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Jacek Lagowski		University of South Florida 4202 Fowler Avenue Tampa, FL 33620	
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13. ABSTRACT (Maximum 200 words)  The program was aimed at advancing the fundamental and applied knowledge required for defect engineering in bulk GaAs. It was focussed on deep level defects, especially antisite defects such as arsenic-antisite related EL2 defect and gallium antisite acceptor defects. Using non-stoichiometric crystal methods, with controlled cooling rates, we established, for the first time, quantitative temperature characteristics of the EL2 formation rate. We determined melt stoichiometry and Fermi energy dependence of the native acceptor concentration. The latter results has stimulated theoretical interest and <i>ab initio</i> calculations by the Xerox group of the negative "U" property of the Ga <sub>As</sub> antisite.  Recently we also completed a study on the hetero-antisite defect, Sb <sub>Ga</sub> . Using our knowledge of EL2 formation/annihilation, we were able to engineer defect structure in arsenic-rich crystal leading to very low EL2 and EL3 concentrations. This enabled the first positive identification of both single and double donor levels of the Sb <sub>Ga</sub> defect otherwise masked by the EL2 and EL3 defects.  The program has also lead to the discovery of a non-contact, no preparation, wafer scale, deep level defect characterization method based on the surface photovoltage transient. The program resulted in eight publications.			
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**FUNDAMENTAL AND APPLIED ASPECTS OF  
DEFECT ENGINEERING IN GaAs**

**FINAL REPORT**

**JACEK LAGOWSKI**

**DECEMBER 08, 1992**

**U.S. ARMY RESEARCH OFFICE**

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## **1.0 FOREWORD**

This work was initiated at The Massachusetts Institute of Technology and it was subsequently transferred to the Center for Microelectronics Research (CMR) at the University of South Florida with the Principal Investigator in late 1989. In view of the eight publications covered by this contract (some of them fairly long), we have included with this report the abstracts of the publications rather than the entire publications. The most recent important results for which reprints are still not available are outlined in Section 3.0 of this report.

A lot of work was also devoted to setting up the experimental facilities of the Defect Engineering Laboratory within the CMR. As a result, the laboratory, supervised by the Principal Investigator, has been equipped with apparatus for growth, annealing and characterization which is especially relevant to realization of the original tasks.

## **2.0 STATEMENT OF THE PROBLEM STUDIED**

The program was aimed at establishing the defect formation characteristics and electronic and optical properties relevant to the engineering of antisite defects in bulk GaAs crystals. It included three inter-related elements:

- engineering of defects via non-stoichiometric crystal growth
- thermal manipulation of defects
- characterization of defects in correlation with theoretical defect models.

Originally focussed on native antisite defects, the program gradually shifted to hetero-antisite defects and especially the  $Sb_{Ga}$  antisite.

## **3.0 SUMMARY OF THE MOST IMPORTANT RESULTS**

Our research was initially focussed on deep level defects, especially antisite defects such as arsenic-antisite related EL2 defect and gallium antisite acceptor defects. Using non-stoichiometric crystal methods, with controlled cooling rates, we established, for the first time, quantitative temperature characteristics of the EL2 formation rate. We determined melt stoichiometry and Fermi energy dependence of the native acceptor concentration. The latter results has stimulated theoretical interest and *ab initio* calculations by the Xerox group of the negative "U" property of the  $Ga_A$ , antisite. The program has also lead to the discovery of a non-contact, no preparation, wafer scale, deep level defect characterization method based on the surface photovoltaic transient.

These results are described in the publications list of Section 4.0 and abstracts are included in Appendix A. Recently we also completed a study on the hetero-antisite defect,  $Sb_{Ga}$ . Using our knowledge of EL2 formation/annihilation, we were able to engineer defect structure in arsenic-rich crystal leading to very low EL2 and EL3 concentrations. This enabled the first positive identification of both single and double donor levels of the  $Sb_{Ga}$  defect otherwise masked by the EL2 and EL3 defects. These results are discussed on the following pages.

### Results on Antimony Hetero-Antisites

We have grown, using the LEC method, a series of GaAs crystals from As-rich melts doped with Sb at a concentration of  $10^{20}$  atoms/cm<sup>3</sup>. The SIMS and PIXE (proton induced x-ray emission) data presented in Table 1 below show the high concentration of Sb atoms in the grown crystals.

Table 1.

#### Impurity Concentration (cm<sup>-3</sup>) In Sb-Doped LEC GaAs

<u>Sample</u>	<u>B</u>	<u>O</u>	<u>Mn</u>	<u>Fe</u>	SIMS Analysis					<u>PIXE</u> Analysis
					<u>Cu</u>	<u>C</u>	<u>Si</u>	<u>Te</u>	<u>Sb</u>	
407	2E17	<9E15	<2E14	1E15	<4E15	2E16	3E15	2E16	2E19	3.0E19
809(s)	6E16	<9E15	<9E13	7E14	<2E15	2E16	4E15	9E16	3E18	2.7E18
856(t)	9E16	<9E15	<1E14	9E14	<3E15	<9E15	3E15	5E17	3E19	6.5E19

(s) -- seed portion of ingot

(t) -- tail portion of ingot

One should expect a fraction of about  $10^{-3}$  of Sb to incorporate on the As-site and, therefore, form a double donor defect with a first ionization level at about 0.48 eV below the conduction band (as previously determined from temperature dependance of the Hall effect and from an optical absorption threshold). To our surprise, we found no new DLTS peak in GaAs:Sb. The DLTS spectrum revealed a 280K peak (larger than usual), but the corresponding trap energy (0.55 eV) was somewhat higher than the suggested 0.48 eV. A very similar position peak is commonly observed in GaAs, it is denoted as the EL3 trap, and it has only recently been identified as an oxygen-related center, possibly due to a negative "U" oxygen-arsenic vacancy complex. Relation to oxygen was unambiguous. It was established using an oxygen LVM signature obtained on crystals which we grew and doped with oxygen isotope <sup>18</sup>O.

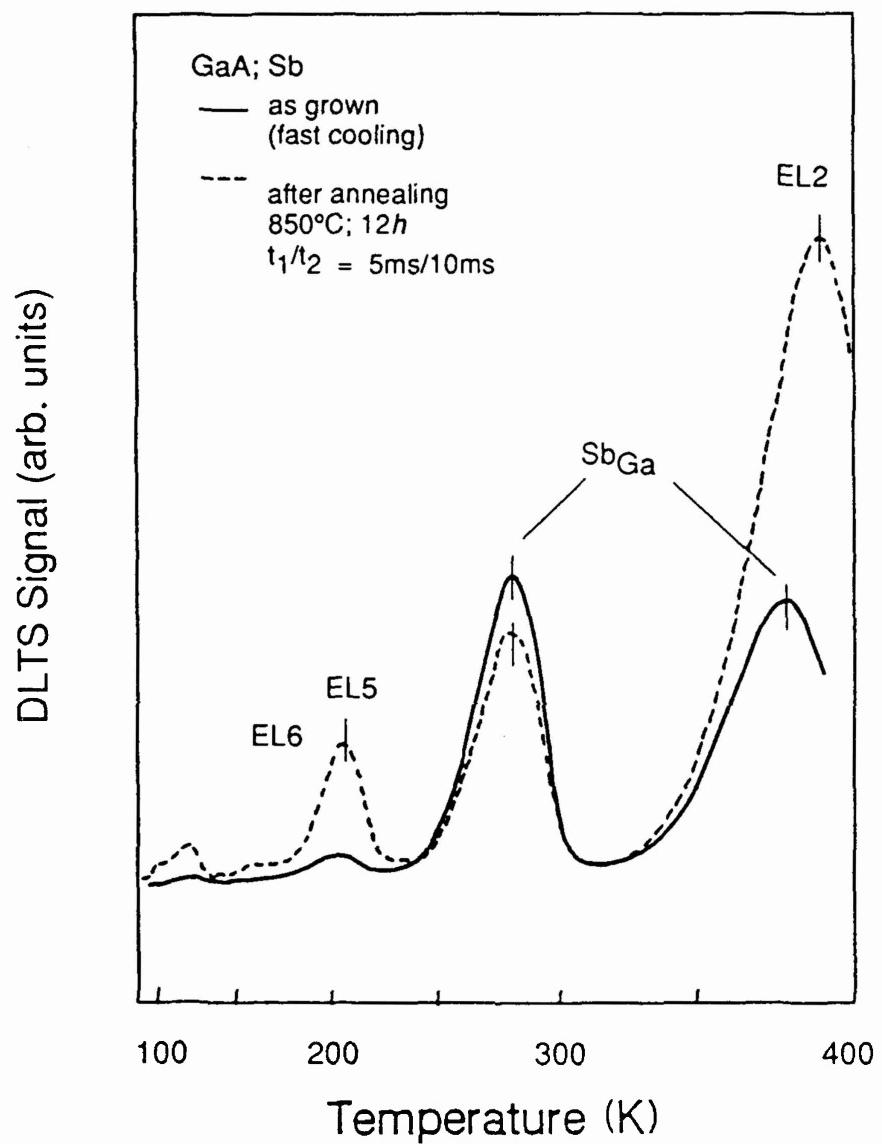
The possibility that  $Sb_{Ga}$  defect has energy level almost the same as that of the oxygen had to be carefully re-examined, especially since the Lund group reported identical  $Sb_{Ga}$  related DLTS peak in epitaxial MOVPE GaAs doped with Sb. To resolve the problem, we grew GaAs:Sb crystals with larger Sb concentrations and we rapidly cooled the crystal after solidification. We expected to eliminate the EL2 defect and to observe the  $Sb_{Ga}$  via levels: the 0.55 eV single donor 0/1+ level in question and the higher energy 1+/2+ level otherwise obscured by the EL2 background. The experiment has fully confirmed these expectations. Two peaks are clearly visible in the DLTS spectrum of Figure 1 for the fast-cooled GaAs:Sb crystal.

The corresponding level's concentration is about  $4 \times 10^{15} \text{ cm}^{-3}$  and it is by a factor of ten larger than the oxygen concentration (determined from LVM oxygen line intensity). The higher temperature peak is only slightly lower in magnitude than the low-temperature one, and consistent with the filling factor effects (depletion region edge effects).

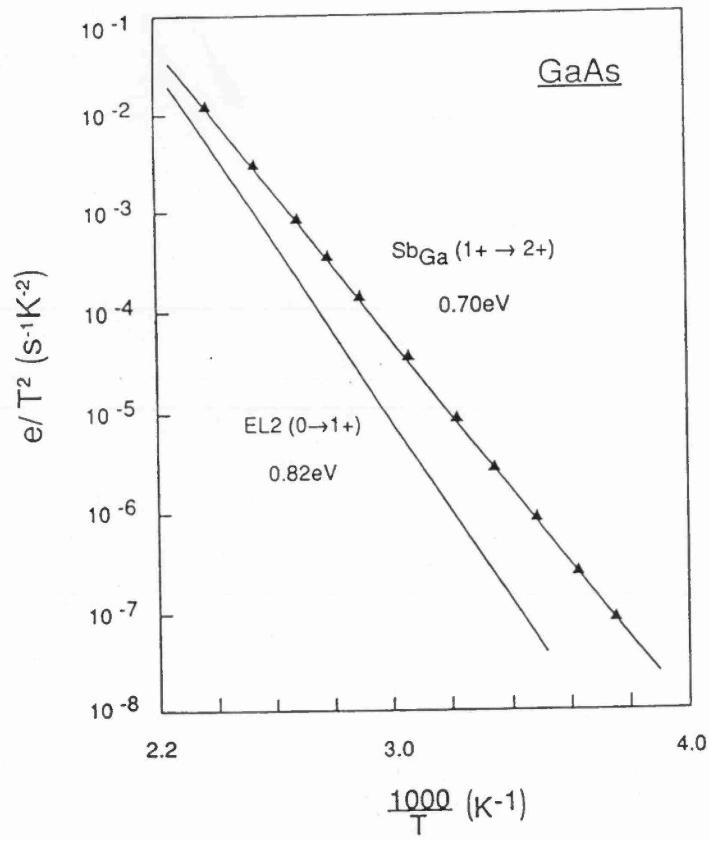
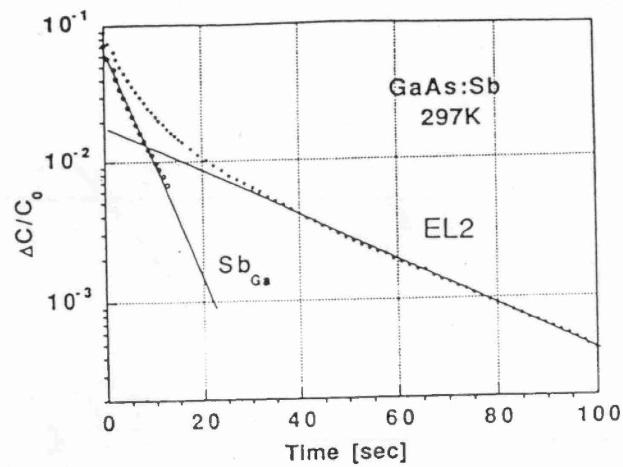
On the same crystal, the capacitance transient in the EL2 time constant range revealed two components (see Figure 2). The slow component corresponds to residual EL2 while the dominant fast component is a new double donor level related to Sb. Figures 2 through 4 give emission rate data,  $e/T^2$ , and corresponding activation energies for Sb-related levels and similar levels in GaAs not related to Sb. Emission rate values are so close that level separation is not possible based on standard DLTS alone. Annihilation of EL2, reliable LVM determination of oxygen content, and direct capacitance transient analysis had to be used together.

### Defects in Ga-rich GaAs

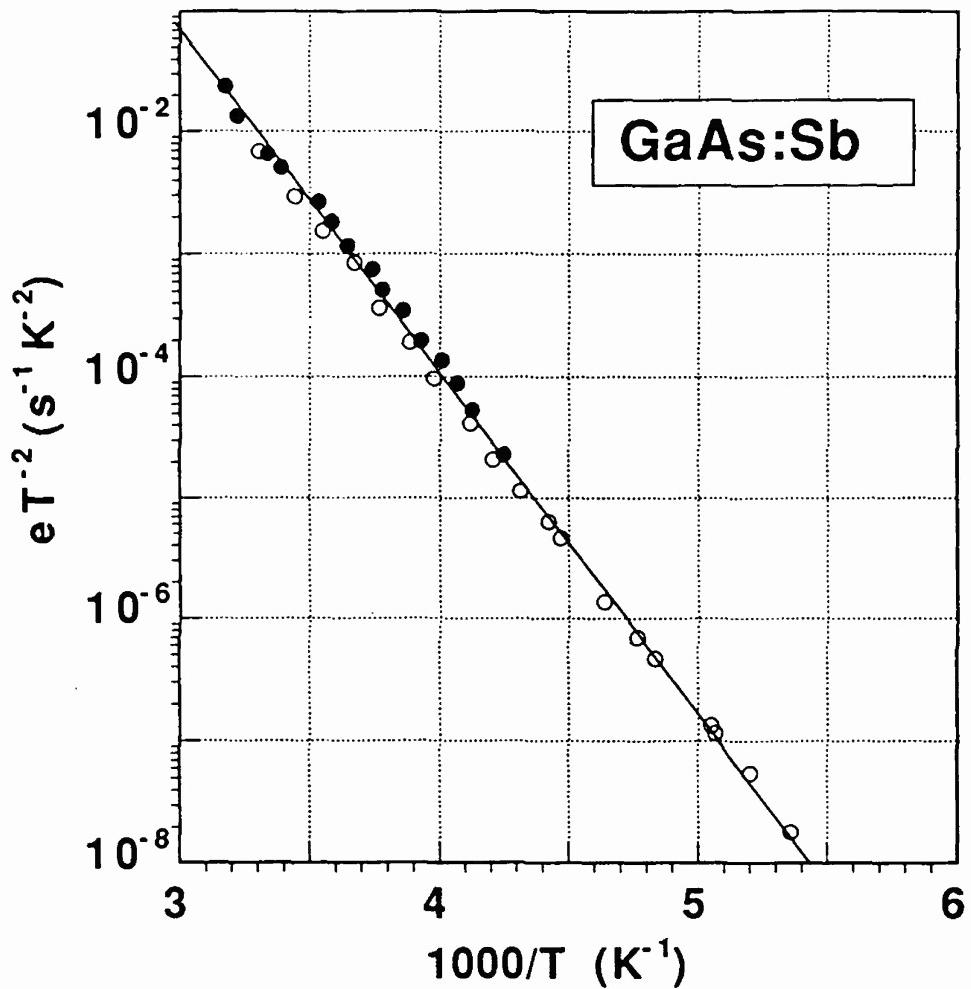
Our results on defects in n- and p-type Ga-rich GaAs has gained theoretical interest and were taken as an indication of the Fermi energy controlled configurational re-arrangement of the cation antisite  $\text{Ga}_{\text{A}s}$ . In Figure 5, such a re-arrangement is manifested by a dominant higher temperature hole-trap DLTS peak in n-type GaAs, while, in p-type material, the low temperature hole-trap dominates. Theoretical calculations by the Xerox group assign such behavior to a negative "U" property of distorted  $\text{Ga}_{\text{A}s}$  antisite in n-type (configuration (a)) and a standard  $\text{Ga}_{\text{A}s}$  atomic arrangement in p-type (configuration (b)). Similar properties should be exhibited by other cation antisites.



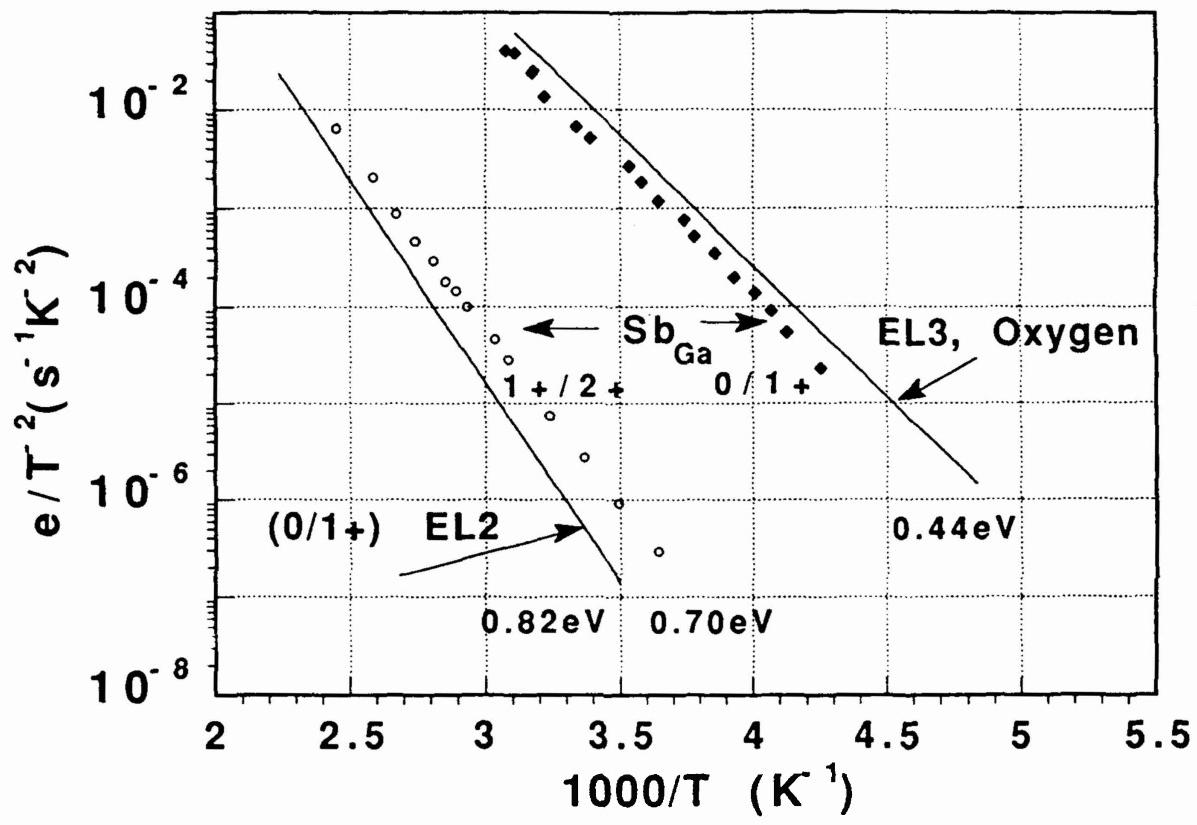
**Figure 1.** DLTS spectra for LEC GaAs doped with antimony and grown under arsenic-rich conditions. Very fast cooling of the "as grown" crystal reduced formation of the EL2, thus exposing the high temperature  $\text{Sb}_{\text{Ga}}$  peak. An 850°C annealing creates the EL2 which completely overshadows the  $\text{Sb}_{\text{Ga}}$  peak.



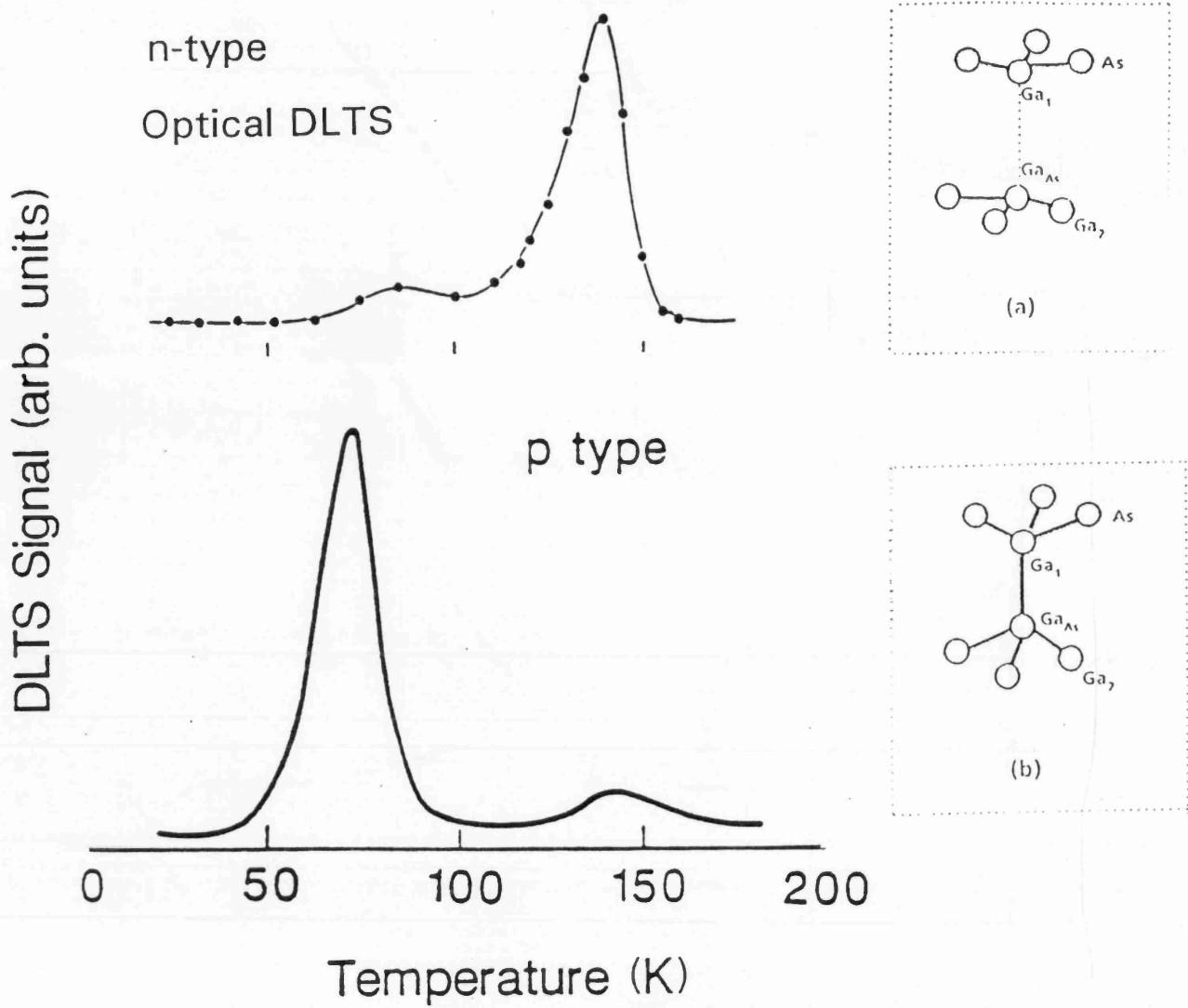
**Figure 2.** Capacitance transients of an LEC GaAs:Sb crystal cooled very fast after solidification. The two transient components correspond to EL2 (slow transient) and Sb<sub>As</sub> (fast transient).



**Figure 3.** Thermal activation plot of electron emission from the single donor state of an  $Sb_{Ga}$  defect measured by us in LEC GaAs:Sb (full circles) and by the Lund group<sup>[47,48]</sup> (open circles) on MOVPE layers.



**Figure 4.** Thermal emission rate data for Sb levels and other similar levels in GaAs.



**Figure 5.** DLTS spectra of hole traps in n-type and p-type LEC GaAs grown under Ga-rich conditions. Note that a high temperature peak dominates in n-type GaAs while shallower defect dominates in p-type GaAs. The *ab initio* pseudopotential calculations of the Xerox groups predicted such behavior for the  $\text{Ga}_{\text{As}}$  antisite defect. Their corresponding atomic defect models are also shown as (a) and (b).

## **4.0 LIST OF PUBLICATIONS AND TECHNICAL REPORTS**

### Defect Properties and Engineering

1. "GaAs vs. Si: Challenges, Hopes and Defect Engineering," J. Lagowski, Acta Physica Polonica, A77, 311 (1990).
2. "Raman Scattering from the Intrinsic 68-meV Acceptor in GaAs," J. Wagner, K.Y. Ko, and J. Lagowski, Phys. Rev., 43(b), 5163 (1991).
3. "Signature of the Gallium-Oxygen-Gallium Defect in GaAs by Deep Level Transient Spectroscopy," S.T. Neild, M. Skowronski, J. Lagowski, Appl. Phys. Lett., 58, 859 (1991).
4. "Optical Characteristics of Single and Double Donor Levels of the EL2 Defect in GaAs," J. Lagowski and M. Matsui, in preparation for publication in J. Appl. Phys.
5. "Electronic Levels of Sb<sub>A</sub> Hetero-antisite Defect in GaAs," J. Lagowski, A. Morawski, P. Edelman, in preparation for publication in Appl. Phys. Lett.

### Defect Diagnostics

6. "Photoluminescence and Minority Carrier Diffusion Length Imaging in Silicon and GaAs," P. Edelman, W. Henley, J. Lagowski, Semicond. Sci. Technol., 7, A22 (1992).
7. "Non-Contact Deep Level Transient Spectroscopy (DLTS) based on Surface Photovoltage," J. Lagowski, A. Morawski, P. Edelman, Semicond. Sci. Technol., 7, A211 (1992).
8. "Non-Contact, No Wafer Preparation Deep Level Transient Spectroscopy Based on Surface Photovoltage," J. Lagowski, A. Morawski, and P. Edelman, Jap. J. Appl. Phys., 31, L1185 (1992).

**5.0 LIST OF PARTICIPATING SCIENTIFIC PERSONNEL**

J. Lagowski -- Principal Investigator  
P. Edelman -- Research Scientist  
A. Morawski -- Research Scientist  
K. Y. Ko -- Postdoctoral Fellow  
W. Henley -- Graduate Student (Doctorate completion expected in the Fall 1993)  
M. Dexter -- Graduate Student (Doctorate completion expected in the Fall 1993 or the Spring 1994)

**6.0 INVENTIONS**

"Non-Contact, Wafer-Scale Surface Photovoltage/Deep Level Transient Spectroscopy"

## GaAs VERSUS Si-HOPES, CHALLENGES AND DEFECT ENGINEERING\*

(INVITED PAPER)

By J. LAGOWSKI

Massachusetts Institute of Technology, Cambridge, Massachusetts 02139, USA

In semiconductor electronic silicon occupies the uncontested "number one" position. "Number two" is taken by GaAs and related compounds. Promoters of military applications tend to place HgCdTe in the "number three" spot. This pattern is not reflected at all by the fundamental research. Here GaAs is clearly number one with no challenger in sight. II-VI compounds are number two. Some experts claim that Si would not reach the TOP 10 list if not for "defect engineering" — a new application-driven trend which has revitalized an interest in fundamental understanding. This new trend is discussed using the examples of metal contamination in Si and native defect-related problems in GaAs.

PACS numbers: 72.20.-i, 72.80.Ey

### *I. Introduction*

There are many ways to compare semiconductor materials and the outcome is often determined by the employed criteria. A solid state physicist would very likely base his reasoning on the energy band structure. Such approach uncovers the overwhelming advantage of GaAs over Si and shows clearly that GaAs can do what Si cannot. From the band structure in Fig. 1 it is apparent that GaAs with its direct energy gap can be an efficient light emitter. On the other hand, the indirect gap of Si makes it unsuitable for light-emitting devices. In GaAs conducting electrons reside in the  $\Gamma$ -point minimum. They are much lighter than the X-point electrons in Si. Thus, in a low electric field GaAs electrons move much faster (6 times at 300 K) than silicon electrons. This translates to higher speed devices. At a high electric field GaAs electrons are transferred from the low effective mass (high mobility)  $\Gamma$  minimum to the large mass (low mobility) L minimum. This leads to a negative resistance, Gunn oscillations, and the generation of microwaves. The Gunn effect is not possible in silicon, whereby the lowest energy conduction band minimum is characterized by a large effective mass. At room temperature the energy gap of GaAs is 1.43 eV as

\* Proc. XVIII International School on Physics of Semiconducting Compounds, Jaszowiec 1989.

## Raman scattering from the intrinsic 68-meV acceptor in Ga-rich GaAs

J. Wagner

*Fraunhofer Institut für Angewandte Festkörperphysik, Tullastrasse 72, D-7800 Freiburg, Federal Republic of Germany*

K. H. Ko and J. Lagowski

*University of South Florida, Tampa, Florida 33620*

(Received 9 July 1990)

Raman spectroscopy with sub-band-gap excitation has revealed electronic transitions at the 68-meV acceptor present in *p*-type doped Ga-rich GaAs. These transitions, which connect the acceptor ground state with shallow bound excited states, have been used to determine the acceptor binding energy spectroscopically to 71.7 meV. From the quantitative correlation of the Raman line intensity with the acceptor concentration measured by deep-level transient spectroscopy the absolute scattering cross section of  $5 \times 10^{-24} \text{ sr}^{-1} \text{ cm}^2$  has been derived.

Native acceptors in bulk GaAs have recently gained considerable interest. The most widely studied one is a double acceptor with its first and second ionization level at 78 and 203 meV above the valence-band edge, respectively.<sup>1-4</sup> In heat-treated GaAs, two intrinsic acceptor levels at 68 and 174 meV have been reported.<sup>5,6</sup> Using magneto-optical and optically detected magneto-resonance techniques an intrinsic acceptor pair with trigonal point-group symmetry has been identified in semi-insulating (SI) and slightly *p*-type GaAs.<sup>7</sup> In *p*-type doped Ga-rich GaAs a native-single-acceptor 68-meV level has been identified with deep-level transient spectroscopy (DLTS) and has been shown to be a dominant intrinsic acceptor.<sup>3</sup> In addition, electron-spin resonance has revealed a number of presumably intrinsic acceptor levels in undoped SI GaAs.<sup>8-10</sup> At present, there is general consensus that with increasing material purity such intrinsic acceptors play an important role in the compensation mechanism responsible for the SI behavior of undoped GaAs.<sup>10</sup>

The purpose of the present work was to study the intrinsic 68-meV acceptor by Raman spectroscopy and to correlate these data with quantitative results of DLTS measurements. Electronic transitions at the 68-meV acceptor have been observed in the Raman spectrum from which the binding energy of this acceptor has been determined spectroscopically. From the comparison of Raman and DLTS data the Raman scattering cross section has been obtained.

The crystals used in this study were grown by the liquid-encapsulated Czochralski (LEC) technique from nonstoichiometric melts intentionally enriched with gallium. They belong to the same set of crystals studied in Ref. 3. The gallium atomic fraction  $\delta$  in the melt ranged from 0.52 to 0.58. The value of  $\delta$  corresponding to different segments of grown ingots was calculated considering the normal freeze process of the binary compound and taking into account the weight loss during the growth of GaAs caused by evaporation.<sup>11</sup> The crystals

were doped with Zn to ensure *p*-type conductivity and the free hole concentration was between  $2.1 \times 10^{16}$  and  $5.1 \times 10^{16} \text{ cm}^{-3}$  (at 300 K), which is well suited for DLTS measurements of hole traps.

Al Schottky diodes were prepared for the DLTS measurements. The diodes were preselected not to exceed a leakage current of  $2 \times 10^{-5} \text{ A/cm}^2$  at 3-V reverse bias. Measurements were performed at 3-V reverse bias using 4-V forward bias filling pulses. Emission-rate data used in the calculation of the activation energy and the capture cross section were obtained from direct measurements of capacitance transients rather than from the positions of DLTS peaks. For each measuring point the temperature was kept constant and the capacitance transient was monitored using signal averaging over typically 200 to 1000 transients. These conditions minimize various real and apparent effects of electric field and junction current on the emission rate.<sup>12</sup>

The Raman spectra were excited with sub-band-gap light at 1064.4 nm using a cw Nd<sup>3+</sup>-doped yttrium aluminum garnet (YAG) laser. The spectra were recorded in backscattering geometry with the samples cooled to 6 K. Further details of the Raman experiment are given in Ref. 13.

The DLTS spectrum shown in Fig. 1 was recorded for a sample with a corresponding Ga atomic fraction of 0.57 (see also Ref. 3). The spectrum was obtained under 3-V reverse bias and the gate setting  $t_1/t_2 = (5 \text{ ms})/(10 \text{ ms})$ . The dominant  $H1^*$  peak (we adopt here the peak notation after Ref. 3) corresponds to a level with the hole emission activation energy of  $68 \pm 5 \text{ meV}$ . The hole emission thermal activation plot for this trap is shown in Fig. 2. The emission rate data in that figure were obtained from capacitance transients measured at different temperatures.

The Raman spectrum from material with a corresponding Ga atomic fraction of 0.57 is displayed in Fig. 3(a). This spectrum is dominated by intrinsic second-order phonon scattering. In Fig. 3(b) the corresponding

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SIGNATURE OF THE OXYGEN-VACANCY DEFECT IN GaAs BY  
DEEP LEVEL TRANSIENT SPECTROSCOPY MEASUREMENTS

S. T. Neild and M. Skowronski

Department of Metallurgical Engineering and Materials Science  
Carnegie Mellon University, Pittsburgh, PA 15213

J. Lagowski

Department of Electrical Engineering  
University of South Florida, Tampa, FL 33620

Experimental results of deep level transient spectroscopy (DLTS) and Fourier transform infrared spectroscopy (FTIR) on oxygen doped GaAs crystals are presented. A positive identification of the DLTS signature of the oxygen-arsenic vacancy ( $O-V_{As}$ ) defect is made through correlation with FTIR measurements. The energy level for the two electron state of the defect is found to be located at 0.55 eV below the conduction band, and direct proof is given that this center is actually a negative-U center. A calibration factor for local vibrational mode (LVM) absorption is calculated to be  $8 \times 10^{16} \text{ cm}^{-1}$ .

Published in *Appl. Phys. Lett.*, **58**, 859 (1991).

# Photoluminescence and minority carrier diffusion length imaging in silicon and GaAs

P Edelman<sup>†</sup>, W Henley and J Lagowski

Center for Microelectronics Research, University of South Florida, Tampa, Florida,  
USA

**Abstract.** The correlation between the distribution of minority carrier diffusion lengths monitored using surface photovoltage measurements and the distribution of photoluminescence intensity has been investigated in silicon and III–V semiconductor materials. Photovoltage measurements have been performed using a non-contact probe and constant photon flux method at room temperature, while the scanning photoluminescence measurements have been done in the room to liquid helium temperature range. To demonstrate the combined capabilities of the two methods, assessments of a liquid phase electroepitaxial  $In_xGa_{1-x}As$  single crystal and 450 °C annealed silicon have been performed. The discussion emphasises the effects of non-uniform carbon distribution in InGaAs and oxygen-related defects in silicon.

## 1. Introduction

The very tight material standards dictated by reliability and performance considerations and the manufacturing yields of electronic devices pose increasing demands on wafer characterisation. The average values of various material parameters, e.g. carrier concentration and mobility, used so far to assess wafer quality are not sufficient. In recent years a strong and growing demand has developed for characterisation techniques capable of giving a clear picture of wafer inhomogeneities before as well as during processing. Such techniques should be non-destructive, i.e. leaving material suitable for further processing. They should also be applicable to standard production line wafers regarding size, thickness and lack of special surface preparation. The final important factor is the speed of measurements.

Photoluminescence (PL) mapping has become a well recognised method used for homogeneity analysis of III–V materials [1, 2]. PL is a fast, non-contact, non-destructive and well established characterisation method widely used for shallow and deep level defect studies. It is well documented too, so interpretation of results can be based on scientifically sound foundations. The often quoted limitations of PL mapping relate to uncertainties brought up by the spatial variation of the non-radiative lifetime  $\tau_{NR}$ .

Surface photovoltage (SPV) spectroscopy is a known method used for determination of diffusion length  $L$  and minority carrier lifetime. A novel SPV approach is based

on using a truly non-contact capacitance-coupled photovoltage probe and a new constant photon flux linear principle [3]. The SPV method also fulfils the requirement for non-contact, fast, wafer-scale measurements dictated by device-processing needs.

Photoluminescence and SPV methods have thus far been used separately and still provide significant information for device fabrication. However, the combination of the two can lead to a better understanding of physical phenomena taking place during growth and processing. If a wafer-imaging technique is to be used for something more than 'accept or reject' testing, the main problem still present is: what is behind the non-homogeneous picture, what is the mechanism of observed contrasts, what is its exact origin and what is its cause? In this work we investigated the origin of contrast on PL images of electroepitaxial  $In_xGa_{1-x}As$  and of CZ silicon wafers after thermal donor activation treatment.

## 2. Experimental details

### 2.1. Samples

The silicon samples used were commercial 4 inch Czochralski (CZ) wafers. They were p-type, approximately  $10^{15} \text{ cm}^{-3}$ , B-doped, (100), of interstitial oxygen concentration nominally  $[O_i] = 15 \text{ ppm}$ . The silicon sample was subjected to 16 h annealing at 450 °C, which produces thermal donors as well as sharp PL lines [4, 5]. The  $In_{0.035}Ga_{0.965}As$  sample was a 1 inch diameter, 3 mm thick (100) single crystal wafer grown electroepitaxially on (100) GaAs substrate [6]. The InGaAs material was

<sup>†</sup> On leave from Institute of Electron Technology, 02-668 Warsaw, Poland.

## NON-CONTACT, NO WAFER PREPARATION DEEP LEVEL TRANSIENT SPECTROSCOPY (DLTS) BASED ON SURFACE PHOTOVOLTAGE (SPV)

JACEK LAGOWSKI, PIOTR EDELMAN,<sup>\*</sup> AND MARK DEXTER

Center for Microelectronics Research, University of South Florida, Tampa, FL

### ABSTRACT

This work reports on the theoretical modeling and experimental investigation of isothermal SPV-DLTS based on the rate window concept. Experimental implementation of the technique is done using computer analysis of the SPV transients after ceasing the illumination. The transient involves two processes -- a recombination of excess minority carriers and a thermal emission of carriers trapped by surface states and bulk defects. The later process is the key one for deep level defect determination.

The upper limit for the measurable deep level emission rate is provided by the recombination lifetime. This limit often exceeds, by orders of magnitude, the standard  $10^3 \text{ s}^{-1}$  limit in capacitance DLTS. The sensitivity of SPV-DLTS is of the same order as that of optical capacitance DLTS.

### INTRODUCTION

Deep Level Transient Spectroscopy, DLTS, is the leading technique for electrical characterization of deep level defects in semiconductors[1,2]. From characteristics values of the electron (hole) emission rates, it enables electronic level identification and offers detection sensitivity as high as  $10^9 \text{ cm}^{-3}$ . In standard DLTS a rectifying test junction, such as a metal-semiconductor Schottky barrier or a p-n junction, is prepared and deep levels are investigated within the junction space charge region. The measurement includes two stages: (1) deep level occupation is disturbed using a junction bias or illumination; and (2) the equilibrium is restored by a thermal emission of electrons (holes) from deep levels to the conduction (valence) band. During the second stage, a transient of any semiconductor property which directly or indirectly senses the deep level occupation can be employed for the emission rate measurements. Junction capacitance transient and junction current transient have been the methods most frequently used. A search for non-contact DLTS has been primarily based on a microwave sensing of the trap occupation[3]. Optical sensing of deep levels, based on non-linear phenomena, has also been investigated[4]. We have only recently proposed deep level sensing using surface photovoltaic (SPV), i.e., the photo-induced change of the surface potential barrier[5]. This phenomenon is the basis for SPV-DLTS.

### RESULTS AND DISCUSSION

Surface photovoltaic measurements can be done in a non-contact, wafer-scale manner without preparation of any junctions or electrical contacts. For

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<sup>\*</sup> On leave from Institute of Electron Technology, 02-668 Warsaw, Poland.

## Non-Contact, No Wafer Preparation Deep Level Transient Spectroscopy Based on Surface Photovoltage

Jacek LAGOWSKI, Andrzej MORAWSKI and Piotr EDELMAN\*

Center for Microelectronics Research, University of South Florida,  
4202 Fowler Avenue, ENG 118 Tampa, FL 33620-5350, USA

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We discuss a novel approach to Deep Level Transient Spectroscopy (DLTS) in which the emission of trapped minority carriers is analyzed employing the surface photovoltage (SPV) transient as measured in a non-contact manner on the native depletion barrier on semiconductor surfaces. Optical excitation is used as the trap-filling pulse. Experiments done on n-type GaAs demonstrate that the SPV-DLTS is suitable for wafer-scale, non-contact determination of deep level defects on semiconductor surfaces. The SPV approach can monitor emission rates up to  $10^6 \text{ s}^{-1}$  which is  $10^2$  to  $10^3$  above the limit of standard capacitance DLTS. The sensitivity of the method is comparable to that of the optical capacitance DLTS.

**KEYWORDS:** semiconductor characterization, deep levels, surface states, non-contact measurement, surface photovoltage

Deep Level Transient Spectroscopy, DLTS, is the leading technique for electrical characterization of deep level defects in semiconductors.<sup>1,2)</sup> From characteristics values of the electron (hole) emission rates, it enables electronic level identification and offers detection sensitivity as high as  $10^9 \text{ cm}^{-3}$ . In standard DLTS a rectifying test junction, such as a metal-semiconductor Schottky barrier or a p-n junction, is prepared and deep levels are investigated within the junction space charge region. The measurement includes two stages: (1) non-equilibrium occupation of deep levels is obtained using a junction bias or illumination; and (2) the equilibrium is restored by a thermal emission of electrons (holes) from deep levels to the conduction (valence) band. During the second stage, a transient of any semiconductor property which directly or indirectly senses the deep level occupation can be employed for the emission rate measurements. Junction capacitance transient and junction current transient have been the methods most frequently used. A search for non-contact DLTS has been primarily based on a microwave sensing of the trap occupation.<sup>3)</sup> Recently, optical sensing of deep levels, based on non-linear phenomena, has also been investigated.<sup>4)</sup>

In this paper we discuss a non-contact version of DLTS which enables wafer-scale measurements without preparation of any junctions or electrical contacts. The non-contact characteristic is achieved by using the surface photovoltage, SPV, transient for monitoring of the electron (hole) emission. The no wafer preparation feature is realized by performing the measurements on native surface barriers rather than on fabricated p-n or M-S junctions.

GaAs wafers, n-type, Te-doped with a free carrier concentration of about  $(5 \text{ to } 8) \times 10^{16} \text{ cm}^{-3}$  at 300 K, were employed in this study. Wafers were chemo-mechanically polished using a Clorox-water solution. The measured

wafer was placed on a chuck which could be heated up to about 100°C. The aluminum chuck served as a ground electrode. Its surface was coated with an electrically isolating aluminum oxide and there was no electrical contact between the wafer and the chuck. The pick-up electrode was placed in the center of a black plastic disk (about 1.5 inch diameter) which blocked background room light from reaching the active wafer area under the probe. The surface photovoltage, i.e., the change of the surface potential barrier under illumination, was generated by 10 mW pulses of He-Neon laser transmitted by a glass fiber bundle coupled to the pick-up electrode. It was measured as a voltage change between the pick-up electrode and the ground using a unity-gain FET preamplifier and a signal averager. A transparent conducting pick-up electrode (indium tin oxide layer deposited on glass) approximately 2.5 mm in diameter was typically placed about 0.2 mm to 0.5 mm above the wafer.<sup>5)</sup> For non-contact SPV measurements, it is important to note that the pick-up electrode, the wafer and the chuck, although not directly connected, are all capacitively coupled. The capacitance pick-up electrode-wafer is the smallest one and, therefore, it dominates in the series of capacitors.

The present study was primarily intended as a quick feasibility test. Thus, the measurements were performed using existing laboratory equipment and data from the signal averager was manually transferred to a personal computer for further processing.

A room-temperature SPV transient is shown in Fig. 1. The "light off" transient contains two distinct components similar to photo-current decay in photoinduced current transient spectroscopy.<sup>6)</sup> The very fast initial decay is associated with recombination of the excess minority carriers, i.e., holes in the present case. The second, much slower, transient component is associated with a release of holes trapped under illumination by the surface states and/or the bulk traps within the surface space charge region. The "trapping" component has an

\*On leave from Institute of Electron Technology, Warsaw PL02668, Poland.